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HIGH STRAIN RATE DAMAGE DEVELOPMENT AND FAILURE MECHANISMS IN TUNGSTEN HEAVY ALLOYS

FINAL REPORT

By

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13. ABSTRACT *(Maximum 200 words)*

A wide range of tungsten heavy alloy microstructures was subjected to high strain rate tensile and compressive loading. It is found that there exists a critical strain for the onset of unstable shear under compressive loading, which is related to the achievement of a viable mean free path for shear band propagation between tungsten particles. The critical strain thus relates to the W-grains, and not the matrix. Similarly, high strain rate tensile strength and toughness is degraded by the inherently weak W-W interfaces in all current alloys. The key to improvement in this case, will be the achievement of matrix "wetting" of these surfaces, and thereby "ductilizing" them.

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I. STATEMENT OF THE PROBLEM

Kinetic energy armor penetration requires optimum material performance under dynamic loading conditions for both tensile and compressive states of stress. One of the two candidate materials for this application is the class of tungsten heavy alloys (WHA), which is a composite material. Damage development in this complex system under the conditions described above is poorly understood. In order to address this issue, an experimental project was undertaken with the objectives of establishing: 1) the role of microstructure in damage initiation and growth, and (2) the principal factors whose optimization would affect the greatest enhancement of WHA ballistic performance.

II. SUMMARY OF IMPORTANT RESULTS

Work on this project has resulted in a number of publications [1-9]. Because of the remarkably different damage mechanisms involved, it seems reasonable to consider separately their contributions to defining the response of WHA to tensile versus compressive states of stress.

A. Compression

One of the most interesting observations was the fact that under rapid loading some tungsten heavy alloys exhibit unstable (possibly adiabatic) shear bands, while others deformed to the same strain level clearly do not [4-9]. On the other hand, their stress-strain curves can be nearly identical (Figure 1), each type showing hardening followed by softening. Factors which seem to enhance the propensity for unstable shear in WHA are [7] (1) prior deformation (usually swaging); (2) the presence of defects (voids); (3) fine grain size, (4) matrix thermomechanical properties which intrinsically favor adiabatic instability nucleation; and (5) possibly matrix precipitation of tungsten particles.

It certainly is not surprising that the intrinsic stability of the material during high strain rate shear should be important. A factor which measures this is the adiabaticity deformation parameter [10]

$$k = \frac{K}{C_p h^2 \dot{\gamma}_o}$$

where K is the thermal conductivity, C_p is the specific heat, h is the gage width of the shear zone, and $\dot{\gamma}_o$ is the initial shear strain rate. For similar values of h and $\dot{\gamma}_o$, it is the ratio K/C_p which describes the first order relative tendency toward unstable shear in different materials; the smaller the ratio, the more likely the possibility of adiabatic shear.

Shown in Table I are K/C_p values for some current (and candidate)* WHA constituents/matrices; several facts are immediately evident. First, current matrices (Ni-Fe and Ni-Co) are much more prone to adiabatic shear than is the principal composite component, tungsten. In fact,

* All must be compatible with the processing requirements inherent in liquid phase sintering.

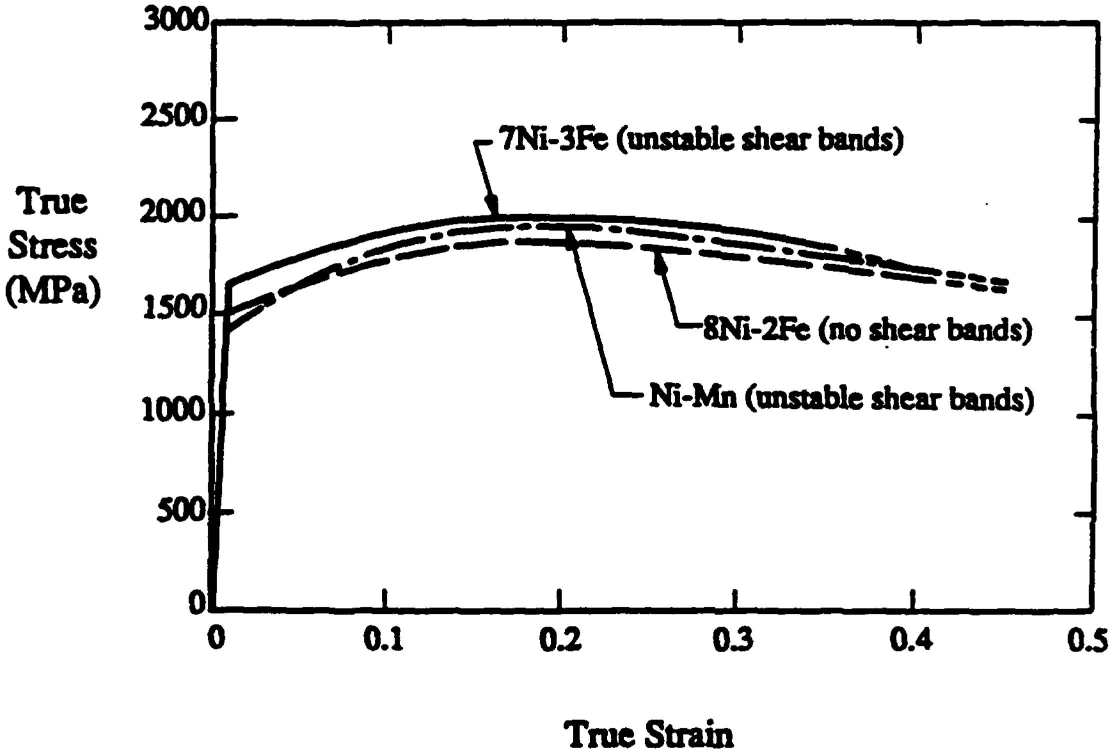


FIGURE 1. Compressive deformation at a dynamic strain rate of approximately 5500 s^{-1} for three alloys; stress-strain curves are virtually identical, but only two exhibit unstable shear.

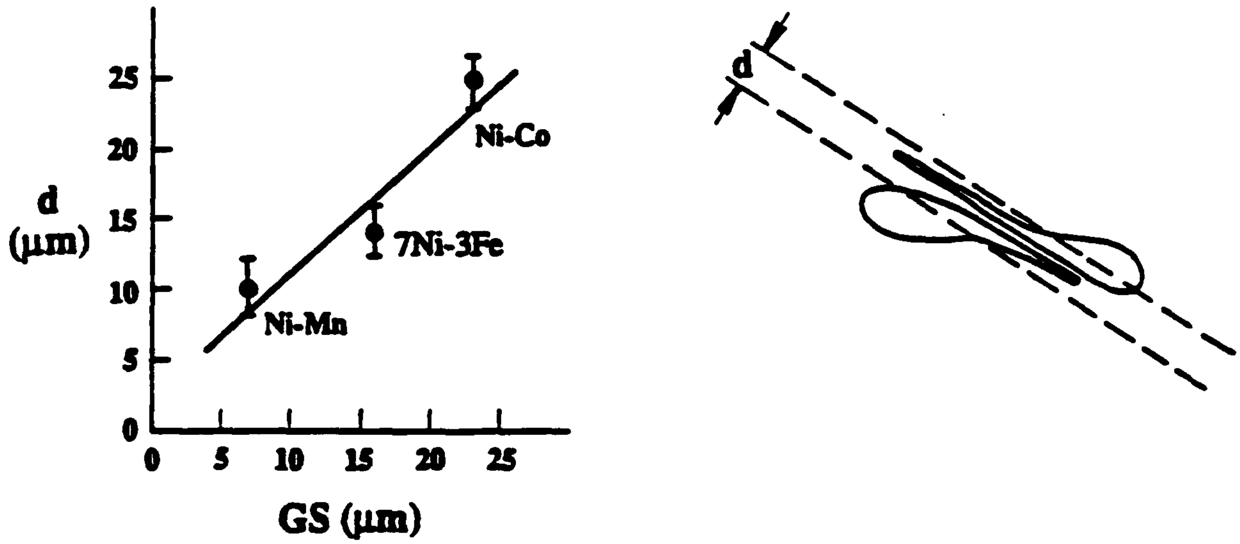
adiabatic shear of pure tungsten has never been documented. On the other hand, the matrices above compare favorably (K/C_p = 0.189 and 0.183) with depleted uranium (K/C_p = 0.239), which is known to shear adiabatically.

Table I.
Adiabatic Shear Parameters

Material	$K(W \cdot m^{-1}K^{-1})$	$C_p(J \cdot kg^{-1}K^{-1})$	K/C
Fe	78	456	0.171
Ni	89	452	0.197
Co	67	380	0.176
NiFe (7:3)	—	—	0.189
NiCo (6:3)	—	—	0.183
Mn	7.8	486	0.016
NiMn (4:6)	—	—	0.088
Hf	22.9	147	0.156
W	174	138	1.26
U	28	117	0.239

Theoretically, the critical strain at the onset of instability also can be computed, although in the context of the composite system, 90% of which is composed of an interconnected tungsten framework, this is an unrealistic exercise. Our experiments, however, suggest that there actually does exist a critical strain for the onset of macroscopic shear instability, and that this threshold is indeed related to the deformation of the tungsten phase. This may seem like an odd statement, since tungsten *per se* clearly is a least likely candidate for adiabatic shear. However, unstable shear apparently cannot occur in these W-rich composites until such time as there is created within the matrix phase a sufficient mean free path, achieved by the deformation and "rotation" of the W-grains. The width of the resulting shear band, in fact, is directly related to the size of the tungsten particles, as defined and shown in Figure 2. The critical macroscopic strain for the onset of unstable shear is on the order of 0.4, which corresponds to local W-particle strains of about 300% within a broad (multigrain) uniform shear band. Beyond the critical strain, W-grain deformation may exceed 1000% within the unstable shear zone (of width "d," Figure 2) which suddenly forms within the initial broad uniform band.

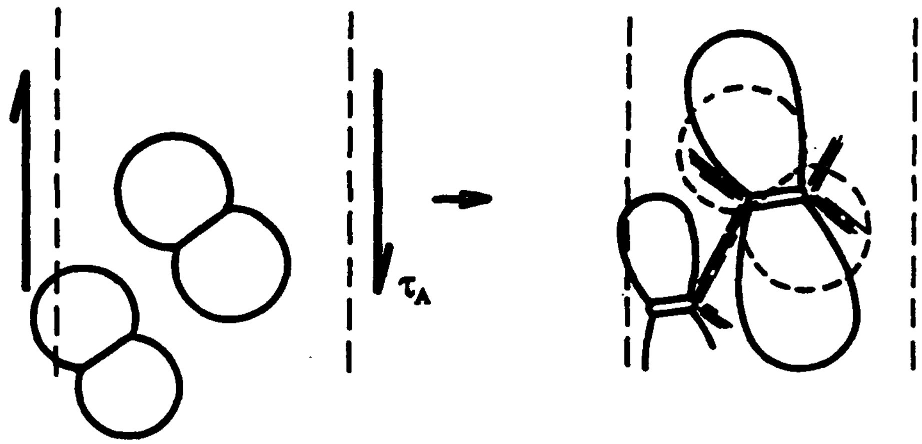
It is important to note that these bands form only in high strain rate compression, and not during equally rapid pure torsion; the basic mechanisms involved are shown in Figure 3. The factor which seems to be responsible for this difference is the presence (in compression) or absence (in pure torsion) of a confining normal force component on the incipient shear plane. During penetration, confinement is present because of the nature of the penetration event. Moreover, Magness and Farrand [11] have shown that the "self-sharpening" of depleted uranium by adiabatic shear band failure near the nose of the penetrator can enhance the penetration performance of that material.



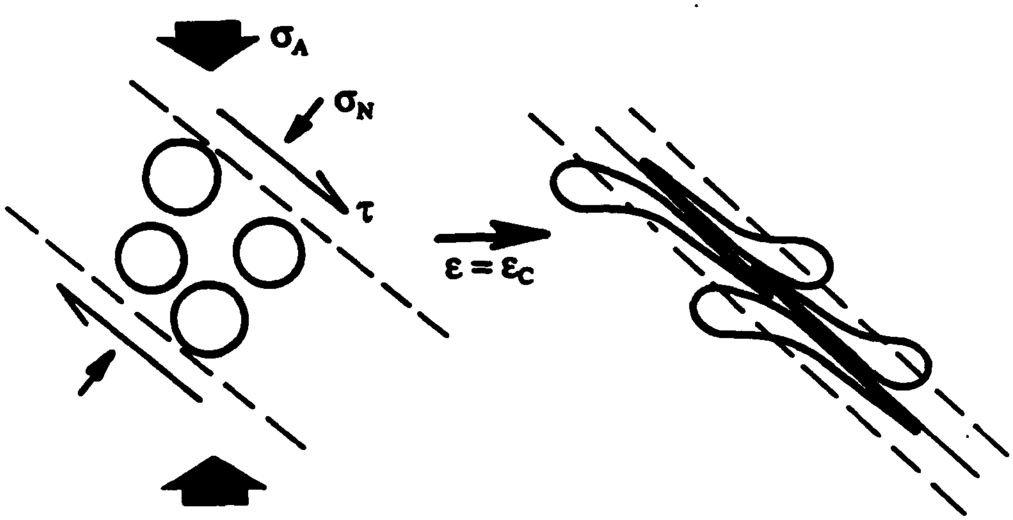
(a) Shear band width versus W-grain size

(b) Definition of shear band width

FIGURE 2. Effect of W-grain size on formation of unstable shear bands.



a) Pure torsion. Applied stress τ_A; constant, multigrain shear bond width, W-grain rotation/elongation, failure by W-W interface microfracture and slip band void nucleation/coalescence.



b) Compression. Applied compressive stress σ_A nucleates multigrain-width stable shear band with shear component τ and normal stress σ_N; at critical strain ε_C, intense unstable (~ single width), narrow band is formed.

FIGURE 3. Shear band development in WHA during high strain rate pure torsion and compression.

B. Tension

The category of tensile loading involves two distinct failure modes, *i.e.*, (1) tensile rupture and (2) fracture. However, we have found [4] that the critical stages of damage development are common in both. For example, at low strain rates, tensile failure (crack nucleation and coalescence) and fracture (sudden propagation of a single pre-existing crack) occur through deformation-induced shear cleavage of W-grains. At high loading rates, on the other hand, tensile failure is caused by the "brittle" nucleation of W-W intergranular cracks, which linkup by matrix void nucleation/coalescence; the same mechanism dominates during dynamic crack extension. Strength rises with increasing strain rate, while elongation (strain to failure) and fracture toughness decrease. The W-W separation mechanism is activated at high strain rates due to the rise in flow stress of tungsten, which eventually exceeds the W-W cohesive strength; this mechanism is activated at surprisingly low strain levels, *i.e.*, before any necking occurs [7].

The influence of the rate-dependent transition between these two mechanisms is reflected in the relative fracture toughness values measured under quasistatic versus dynamic loading conditions. By means of a unique coupled pressure bar technique [12, 13], stress intensity rates in excess of 10^6 MPa $\sqrt{m}\ s^{-1}$ can be obtained. Average experimental values for two tungsten heavy alloys are shown in Table II for both quasistatic and dynamic loading conditions. For both materials, K_{Ic} is reduced at the higher strain rate, markedly so for the Ni-Fe variant. Most interestingly, it has proved possible to model the fracture process theoretically [2, 4] by incorporating the relevant microstructural parameters into an appropriate crack tip failure criterion.

Table II.
Experimental and Theoretical Values for Static and Dynamic Fracture Toughness of WHA

Alloy	Stress Intensity Rate (MPa $\sqrt{m}\ s^{-1}$)	K_{Ic}^{Th} (MPa $\sqrt{m}$)	K_{Ic}^{Exp} (MPa $\sqrt{m}$)
90W-7Ni-3Fe	1	65	71
90W-7Ni-3Fe	10^6	20	28
90W-6Ni-3Co	1	47	46
90W-6Ni-3Co	10^6	40	35

For the quasistatic case, it was observed that within the blunted crack tip plastic zone, initial fairly uniform deformation within both tungsten and matrix is interrupted at a critical local strain ε_C by multiple deformation-induced cracking of W-grains (Figure 4). The crack tip stress field is transferred to the remaining matrix ligaments, which then suffer disbonding from the adjacent tungsten particles and assume the configuration of miniature, reduced section tensile specimens. Their failure by void nucleation and coalescence constitutes the final (local) fracture step. The toughness was calculated using Rice and Johnson's continuum model [14], which assumes that crack extension occurs at a blunted crack tip when critical strains are achieved over a critical microstructural distance, which was chosen to be equal to the average spacing between cracked tungsten particles. The excellent agreement between the experimental and calculated quasistatic toughnesses supports the crack growth assumptions, *i.e.*, void growth and coalescence

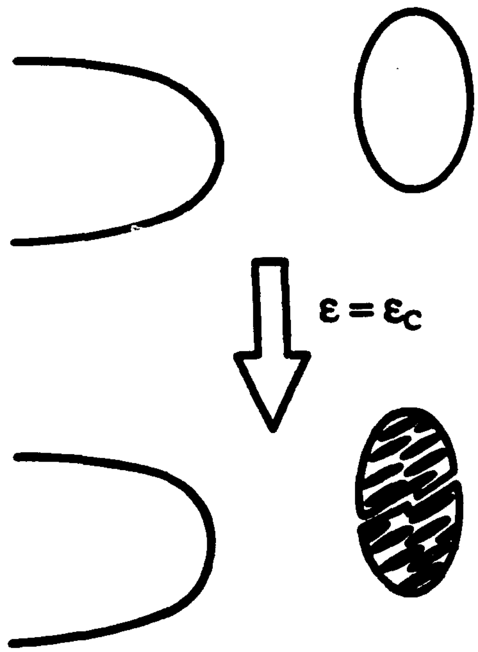


FIGURE 4. Incremental crack tip extension under quasistatic loading; crack tip plasticity permits blunting and deformation-nucleated micro-fracture in nearby W-grains $\varepsilon = \varepsilon_c$; remaining ligament(s) fail by micro-void nucleation/coalescence.

over a region the size of a tungsten particle, and also the implicit assumption that continuum models can be adapted to make useful computations at local regions within large-grained composite systems like the tungsten heavy alloys.

On the other hand, dynamic fracture specimens were found to fail at measured crack opening displacements theoretically characteristic of brittle fracture. Under dynamic loading conditions, the fracture process was dominated by apparent low strain, pure tensile cleavage at W-W interfaces. Toughnesses therefore were calculated using the cleavage model of Ritchie, Knott, and Rice [15], based on the postulate that fracture initiation (Figure 5) occurs when the maximum principal stress at the crack tip equals or exceeds the cleavage stress σ_C (Griffith stress) over some microstructural distance, combined with Tracey's computations [16] of the crack tip behavior under small scale yielding for a rigid plastic material. For both microstructures the critical distance over which the cleavage stress is applied was taken to be the mean spacing between failed tungsten/tungsten interfaces. Again, theoretical and experimental agreement were consistent (Table II).

Clearly, W-W interface microfracture is the weak link in WHA tensile mode failure under dynamic loading. In an effort to explore ways of dealing with this mechanism, heat treatments involving extended sintering times and higher processing temperatures were implemented using a classic 90W-8Ni-2Fe alloy [7]. While the elevated temperature process was ineffectual, producing marked embrittlement, extended sintering times of up to four times normal at the standard temperature produced, under high strain rate conditions, near 50% increases in ductility plus higher hardening rates, although with some loss in overall strength. Microscopic fracture surface examination reveals that the additional time at temperature produces wetting of both surfaces of some (but not all) of the W-W interfaces, which then fail by multivoid nucleation and subsequent fracture of resulting tensile microligaments. The basis for this beneficial effect is shown schematically, versus unmodified material, in Figure 6; clearly this energy absorbing process would be optimized if all W-W interfaces could be "ductilized" by wetting.

C. Summary of Critical Issues

Based on the foregoing, two overarching areas of concern emerge as bases for understanding of, and ability to modify, high strain rate damage development in WHA. On the compressive side, the issue centers about those factors that are, or may be, pertinent to the development of unstable matrix shear band formation, in particular, how to deal with the blocking presence of a high volume fraction of large, stable tungsten particles. For tensile loading conditions, the overriding issue involves control of the W-W interface, the microstructural weak link under dynamic loading for both uncracked and precracked situations.

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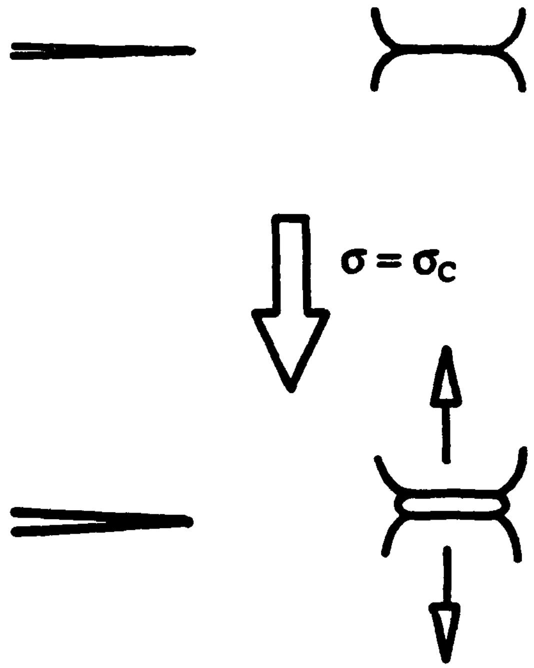
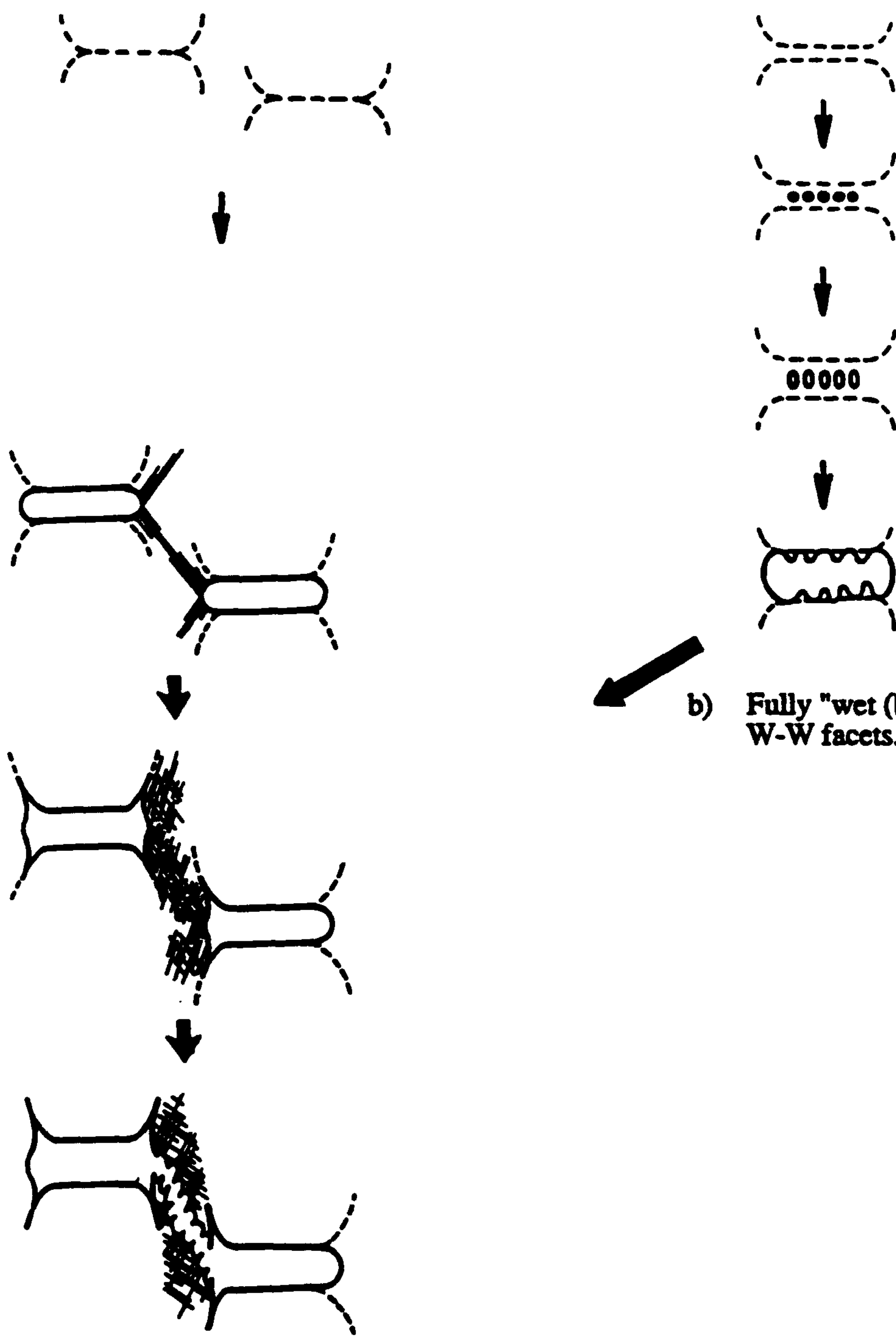


FIGURE 5. Incremental crack tip extension under dynamic loading; little crack tip blunting causes critical stress-nucleated microfracture of nearby W-W interfaces at $\sigma = \sigma_C$; remaining ligaments fail by microvoid nucleation and coalescence.



b) Fully "wet (by matrix) W-W facets.

a) Non-"wet" W-W facets.

FIGURE 6. Influence of W-W interface conditions on high strain rate failure.

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V. TECHNICAL REPORTS

None.

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